Effect of Boron Addition on Structure and Properties of Low Carbon Bainitic Steels

X. M. WANG and X. L. HE

Dept. Materials Physics, University of Science and Technology Beijing, Beijing 100083 China

The effect of trace boron in steels, especially the influence of boron on microstructure and properties of low carbon bainitic steels were summarized. It was pointed out that the hardenability of boron steels is determined by non-equilibrium boron segregation on grain boundaries. The synergistic effect of boron with other elements can enhance the boron effect. The abnormal boron segregation on the moving new grain boundaries is the basic reason of the boron retarding the recrystallization of deformed austenite, the addition of copper and niobium will increase this effect. The addition of boron accelerates the strain induced niobium carbonitride precipitation, and trace boron can refine the microstructure and improve the mechanical properties of HAZ. The mechanisms of boron effect were also expounded preliminarily in this paper.

KEY WORDS: Boron; strain induced precipitation; hardenability; deformed austenite recrystallization.

1. Introduction:

The boron steel had long history before and the researchers had been paying more attention to the boron effect in steels for a long time. A series of good summaries and international symposiums can be accessible¹⁻⁶⁾. Due to the development of modern metallurgical technology, the technique of controlling trace boron tends to be mature. Besides it is found that the boron bearing low carbon bainitic steels have excellent combination of high strength, toughness and field weldability, boron steels have been extensively applied to the engineering machinery, sea platforms and oil pipelines fabrication⁷⁻¹¹⁾. In addition, the Cu-B low carbon bainitic steels have good corrosion resistant yet, so it was widely used in the construction of warships, bridges and sea frameworks¹²⁻¹⁵⁾. Considering energy saving, cost reducing and properties improving, the accelerated cooling and direct quenching technique was widespreadly employed to the controlling rolling and TMP processing, therefore the boron steels gained quick development. The effect of boron on hardenability, austenite recrystallization and strain-induced precipitation arouse the recognition of researchers. This article aims to summarize the effect of trace boron in steels especially the influence of boron on microstructure and properties of low carbon bainitic steels.

2. Experimental materials and methods

The materials involved in this paper were mainly Mn-Nb-B and Cu-Nb-B low carbon bainitic steels, containing 0.02-0.05% carbon. The C-B steel (0.54C-0.75Mn-0.0027B), Fe-30%Ni and Fe-3%Si alloys were employed respectively to study the hardenability mechanism of boron and the behaviors of boron during recrystallization at high temperature. The latter was chosen to prevent the influence of phase transformation during cooling. Furthermore, the boron effect in the low carbon bainite/martensite Cu-B-Nb steel (1[#]), Cu-Nb steel (2[#]) and Nb-B steel (3[#]), was also discussed. All of the thermal simulation tests such as the determination of CCT and PTT curves and the welding simulation were carried on a Gleeble-1500 thermo-simulator. The isothermal phase transformation was conducted in the tin-bath furnace. The distribution of boron was revealed with the aid of particles tracking autoradiography (PTA), the detecting boron sensitivity of this method is 1ppm and the spatial resolution is 2 μm .

3. The effect of boron in steels

3.1 the effect of boron on hardenability and phase transformation of steels

The boron steel has a long history, in the early stages the

boron steels mainly were the quenched and tempered steels with martensite microstructure. Most of researchers were interested in the boron hardenability effect. It was found that the trace boron would enhance the hardenability of steel markedly and this effect had a close relation to the chemical component of the steel and the inclusion elements such as oxygen and nitrogen. When the composition of steels was given the variation of

Fig. 1. half-martensite Jominy distance varied with quenching temperature.

hardenability with the quenching temperature had a peak feature. The thermal history of steel before quenching had obvious influence on the hardenability. There are mainly four explanations on the boron effect mechanism. The first one is that the boron segregation to austenite grain boundaries reduces the grain boundary energy and reduces the preferential nucleation sites for ferrite. The second one is that the boron reduces the self-diffusion coefficient of iron at grain boundaries and reduces the nucleation rate of ferrite. Thirdly, since the grain boundaries is a preferential nucleation sites for ferrite, when boron segregates to grain boundaries, these preferential sites will vanish. Lastly, the fine boride forms along the boundaries and is coherent with the matrix, in this case the ferrite is hard to nucleate at the interface between boride and matrix. It is mutual in all of these explanations that the equilibrium boron segregation at austenite grain boundaries influences the nucleation process of ferrite but it does not affect the thermodynamic characteristics of austenite or ferrite. With respect to the peak feature, Ohmori¹⁶⁾ drew a conclusion that the hardenability of steel reached the highest only if the equilibrium segregation reached the maximum value and the boride was absence.

However, in eighties the comparison of hardenability test results with the PTA photos revealing the distribution of boron in the steel showed that the difference of non-equilibrium segregation state of boron during quenching was the main reason of hardenability variation, so the hardenability not only has a relation to the boron content in steel but also the practical heat treatment processing. That is to say the hardenability will be the highest, if the velocity of boron non-equilibrium segregation formation can be comparable with the incubation of austenite decomposition. If the segregating develops too quickly or too slowly during cooling process the boron effect on hardenability will not be the best¹⁷⁾. For example, the variation curve of half-martensite jominy distance of samples which have same austenite grain size with different quenching temperature is shown in Fig.1, it can be seen that this curve is of peak feature and the peak

temperature is 900°C. For the same samples the PTA photos (Fig.2) indicate that when the steel are quenched from over 900°C (900–1200°C), the boron segregates to the boundary too quickly and the segregation developed into precipitates, causing the boron effect decreases with temperature increases. When steels are quenched lower than 900°C boron segregates too slowly and the boron effect will not be the best, the boron effect declines in the same way.

Fig. 2. The boron distributions at 10.5cm jominy distance of samples quenched at (a) 800°C, (b) 900°C, (c) 1000°C, (d) 1150°C.

In recent years the microalloying steel tends to have higher strength. Considering the weldability the carbon content of steels declines remarkably. Therefore to gain the good synergistic properties, in general the microstructure should be the bainite/ martensite. For increasing the hardenability, the boron effect in this kind of steel will be noticeable. Tomaya Roeski¹⁸⁾ studied the low carbon Cu, Cr steel and found that the hardenability of the steel varies with the holding time after final rolling, When held at short time, the boron cannot segregate to grain boundary and the hardenability declines. Increasing holding time, the boron will segregate to the austenite grain boundaries and the boron precipitates will appear in succession. When obvious boron segregation is present but the boron precipitates does not appear at the austenite grain boundary and the deformation bands, the hardenability of steels will be the best. When there is large amount of boride at the boundary, the hardenability of steel will decline dramatically.

As the Nb, Ti are usually added into the microalloying steel, the study on the synergistic effect of boron with other alloy elements increases markedly. In 1987 Tamehiro³⁾ studied the effect of B, Nb, Ti on austenite decomposition and indicated that there was synergistic effect of Nb+B, Ti+B on retarding $\gamma \rightarrow \alpha$ transformation. They could strongly suppress the formation of ferrite during cooling and the transformation start temperature was decreased to 680–650°C.

The result of CCT curve¹⁹⁾ of Cu-Nb-B bainitic steel

(Fig. 3) shows that the steel has good hardenability, no proectoid ferrite formed when steel is cooled at 2°C/s. The bainitic transformation temperature decreases to 600°C. It appears that the Cu-B has synergistic effect on the phase transformation. As comparable with the result of Tamehiro³⁾, the synergistic effect of Cu-B is bigger than Nb-B.

Fig. 3. CCT Curve of Cu-Nb-B steel

The bainite transformation temperature cooled at 2°C/s of Low Carbon 1[#] and 2[#] copper bearing steels is shown in Table 1²⁰⁾. It can be seen from table 1 that the boron retards bainite transformation effectively. The bainite transformation temperature of steel 1[#] decreases about 20°C. This result indicated that when the boron is added into this kind of steel, the hardenability can be improved markedly and the part of lath bainite, which has better combination of strength and toughness, will increase so the good synergistic properties will be obtained.

Table 1. Transformation temperature of 1[#] and 2[#] steel

No.	B _s (°C)	B _f (°C)
1 [#] (Cu-Nb-B)	627	489
2 [#] (Cu-Nb)	648	526

3.2 Influence of boron on austenite recrystallization

Boron has distinct effect during TMCP of low alloy steel. Firstly the boron can inhibit the deformation austenite recrystallization remarkably. For example the softening curves determined in the low carbon steels which single or combination of Niobium and boron are added in and had been deformed 25% at 1000°C and 900°C are illustrated in Fig. 4²¹⁾. The figure shows the softening process in the plain steel and Nb-free boron steel proceeds much more quickly. Thus the single addition of boron can retard austenite recrystallization but the effect is not appreciable at high temperature. However, due to the strong synergistic affection of niobium and boron, the combined addition of niobium and boron can retard austenite recrystallization markedly at 1000 °C. The variation of boron distribution revealed by PTA on samples deformed at 1000°C and held for different times before quenching in ice brine is shown in Fig.5. It indicates that there is an abnormal boron segregation along some autenite grain boundaries during recrystallization after deformation, which appears as some curly arc lines when the recrystallization starts. This segregation reaches the obvousest when samples are held for 100s then it vanishes gradually with the holding time increasing further. As comparable with the recrystallization curve (Fig.4) it can be found that this kind of segregation mainly appears at the boundaries between the new growing

grain and deformed original grain during recrystallization. This segregation will persist and disappear until the new grains impinges and their motion is halted. Due to the solute drag force, the migration of the new grains is hindered so the recrystllization is hindered. The strong synergistic effect of niobium and boron maybe result from the formation of (Nb, B) complexes. The presence of such complexes at the new boundaries increases the solute drag force of the niobium atoms, decreases the boundary moving velocity and slows down the recrystallization kinetics of steels containing niobium and boron.

Fig. 4. The softening behaviour of the plain C, C+B, Nb and Nb+B steels deformed 0.25 at 1000 and 900°C.

Fig. 5. The boron distribution revealed by PTA in the Nb+B steel held for (a) 3s, (b) 10s, (c) 10s, (d) 100s, (e) 150s, (f) 600s after 25% deformation at 1000°C.

Fig. 6. Softening curve and recrystallization start time, 50% time and stop time of Mn-Nb-B steel

During the development of Mn-Nb-B pipeline steels it has been found that the recrystallization stop temperature of steels with boron is about 950°C. Typical softening curve and recrystallization start time, 50% time and stop time is illustrated in Fig. 6. The recrystallization curve of Cu-Nb-B steels is shown in Fig. 7, it can also be concluded that the combined addition of copper, niobium and boron lead to the recrystallization meet with obstruction after 20% deformation at 950°C. With the increase of copper, the turning point in Fig. 7 moves to left side²²⁾.

Fig. 7. Softening curve of Cu-Nb-B steel

In order to gain a clear idea of behavior of boron in the high temperature recrystallization, a single-phase Fe-30%Ni (FCC) alloy and a Fe-3%Si alloy (BCC) were applied to prevent the influence of phase transformation during cooling^{23,24)}. Figure 8 shows the distribution of boron and the corresponding optical micrograph of the Fe-3%Si alloy which was held 10s at 1000°C after 20% deformation. It is clearly shown that there is no obvious boron segregation on original boundaries (SGB in Fig. 8) but an obvious boron segregation appears along the moving boundaries (MGB in Fig. 8) of the newborn recrystallizing grains. The PTA photo and corresponding metallograph of

Fe-30% Ni alloy, which was held 3s at 1000°C after 20% deformation is given in Fig. 9^{19,24)}, it indicates that at that time the recrystallization was just beginning, the new grain was formed at the triple points and growing. A clear boron segregation is observed on the moving boundaries but there is a very few of segregation on the original boundaries. All these results indicate that when the newborn grains are growing, a boron segregation only exists and persists on these new moving boundaries, as the recrystallization is ended and the moving velocity decreases, this kind of boron segregation begins to decrease and finally disappears. This type of abnormal boron segregation on moving boundary has also been observed in the bainitic steels¹⁷⁾. Two kinds of possible mechanisms were proposed to explain the solute abnormal segregation mentioned above.

Fig. 8. (a) The boron distribution and (b) metallograph of Fe-3% Si sample, the pictures has a mirror relation

Fig. 9. (a) The boron distribution and (b) optical micrograph of Fe-30% Ni sample

the first one is the grain boundary widening mechanisms based on interactions between dislocations and moving boundaries during recrystallization. During the recrystallization process, the new boundaries will move toward the high dislocation density areas, and leave regions with low dislocation density areas, a large number of dislocations will be annihilated in the moving boundaries during recrystallization. Due to the annihilation of dislocations in grain boundaries needing an appreciable time, the external dislocations incorporated in boundary will give the boundary an extra distortion area and the boundary thickness will increase before dislocations disappear. Since the practical width of moving new boundary increases, the area which boron can segregate to enlarges and the amount of segregation atoms increases. Another mechanism said that a high-density dislocation

boundary thickness will increase before dislocations disappear. Since the practical width of moving new boundary increases, the area which boron can segregate to enlarges and the amount of segregation atoms increases. Another mechanism said that a high-density dislocation area which is formed in front of the moving new grains also could adsorb the solute atoms as grain boundaries themselves and this results in the enormous increase in the abnormal segregation of boron on the moving boundaries.

3.3 Effect of boron on strain-induced precipitation

Boron has appreciable effect on the strain-induced precipitation behavior of low carbon steels with Nb, Ti. The strain induced precipitation of C-Mn low carbon bainitic steels with Nb, Ti was studied by Djahazi²⁵⁾. The result show that in Ti steel the precipitates is mainly TiN and the addition of boron does not seem to influence the morphology and distribution of TiN. In Nb-B steels the addition of boron leads the PTT curves shift to left and the nose temperature rises about 15°C, but it seems that boron has little impact on the P_f time. The precipitates are detected as (Nb, Ti)(C, N, B) by EELS, so the boron addition in the steels means increasing the interstitial atom density, and accelerating the precipitate nucleation, and a finer and more dense precipitates will appears. According to the statistic results of precipitates size the presence of boron will increase the diffusion coefficient of niobium in steel.

Yang²⁶⁾ investigated the strain-induced precipitation of Mn-Nb-B steels which bearing different C, B content deformed 20% at four temperatures, the PTT curve of Nb(C, N) is shown in Fig. 10. This PTT curve is of C shape and the nose temperature is 850°C. The result coincides with the work of Djahazi and indicates when the boron is added in steels, the Nb(C,N) precipitation is accelerated and the amount of precipitation increase, the strength of steels is improved too.

The influence of combined addition of boron and other elements on the Nb(CN) precipitation has been studied. The strain relaxing curves of low carbon bainitic steels with Cu (1[#], 2[#] and copper free 3[#] steels) were determined on Gleeble-1500 thermo-simulator, the PTT curves of these steels are shown in Fig. 11²⁷⁾.

From the Fig. 11 the addition of copper and boron leads the PTT curves shift to left obviously and the nose temperature rises from 800 °C to 850°C. The boron effect on Nb(C, N) precipitation is enhanced by the copper addition. It seems that the copper and boron have synergistic effect evidently. These results indicate that the boron addition corresponds to increasing the interstitial atom density so the driving force of nucleation is increased and P_s time is decreased. The boron addition increases the diffusion coefficient of Nb atom and decreases the P_s in the same way. On the other hand the boron atom can segregate to the dislocation and grain boundaries, decreasing the interface energy between the precipitation and matrix therefore the nucleation energy is reduced and the nucleation is accelerated. The reason for the synergistic

function of boron and other elements (like copper) needs further clarification.

Fig. 10. PTT curve of Mn-Nb-B steel

Fig. 11. PTT curves of 1[#], 2[#] and 3[#] steels

3.4 Effect of boron on microstructure and properties of low carbon bainitic steels

3.4.1 The stability of microstructure during tempering

Boron has an obvious influence on the development of microstructure of low carbon bainitic steels during tempering. Figure 12 shows the microstructures of 1[#] steel after quenched and tempered. The micrograph of quenched sample is shown in Fig.12(a), and Figs.12(b) and 12(c) show the samples tempered at 520°C for 30 and 180 minutes. Fig.12(d) gives the metallograph of sample tempered at 600°C for 120 minutes. As can be seen from Fig.12(a) the quenched microstructure of 1[#] steel is mainly composed of lath bainite and a little granular bainite, the proportion of lath bainite is about 75% and the rest is granular bainite. The boundary of granular bainite block is irregular, but in the lath bainite the lath boundary is continuous and parallels each other, but a few of lath boundaries are curly a little. The lath is finer and longer, the width of the lath is relatively big. The Figs. 12(b) and 12(c) show that after 520°C tempering the microstructure are still mainly bainite. With increasing the tempering time the lath boundaries became more and more blurer and discontinuous, the width of laths increases and a part of outlines of lath disappear. As the tempering time increase further the lath width increases more over, the lath boundaries disappear at last. When sample is tempered for 120 minutes at 600°C a majority of the outlines of lath vanished and the retained austenite particle distributed in disorder, the matrix microstructure is mainly granular

Fig. 12. Microstructure of 1[#] samples tempered at 520°C

Fig. 13. Microstructure of 2[#] samples tempered at 520°C

Figure 13 shows the microstructures of quenched and tempered samples of 2[#] steels (boron free). The micrograph of quenched sample is shown in Fig.13(a), and the Figs. 13(b-c) show the results of samples tempered at 520°C for 30 and 180 minutes. As can be seen from Fig.13, the quenched microstructure of 2[#] steels is still mainly composed of lath bainite, but the proportion of lath bainite is about 50% and much less than in 1[#] steel. The laths are coarse, the width is bigger and the parallelism of lath is not good. Comparing the Figs.12(b), 12(c), 13(b) and 13(c), it can be found that when 2[#] steel is tempered at 520°C, the tempering process of the lath microstructure can still be characterized by the bainite lath coalescing, widening and disappearing at last, but the tempering rate is more quickly. Figure 13(d) gives the metallograph of sample tempered at 600°C for 120 minutes, it is basically a quasi-massive microstructure. The results above indicate that boron has marked affection on the evolution of microstructure during tempering. When the boron is absent in steels the quenched microstructure is relatively coarse, the lath width is bigger and their parallelism becomes worse. Meanwhile the

tempering process of lath bainite progressed more quickly. It seems that boron can not only increase hardenability of steels and decrease B_s temperature of steels but also increase the lath bainite proportion, refine the microstructure and makes the lath become more stable, retards the tempering of bainite microstructure.

3.4.2 The properties difference after similar heat treatment

The boron addition increases the amount of lath bainite, and improves the strength of as rolled plates.

Table 2 gives the mechanical properties difference between 1[#] and 2[#] steel after similar 910°C quenched and tempered for 1.5h at 650°C.

Table 2. Mechanical properties of 1[#] and 2[#] steels after 910°C quenched and tempered for 1.5h at 650°C

No.	σ_s (MPa)	σ_b (MPa)	δ_s (%)	ψ (%)	$A_{KV}(J)$ (-40°C)
1	743	785	20	73	116
2	695	755	20	75	118

Table 2 shows these two steels can keep quite high strength, the strength of boron steel (1[#]) is 40 MPa higher than boron free steel (2[#]). At the same time two steels can keep quite high toughness, the impact energy at -40°C is over 100J and the elongation can reach 20%. It seems that the boron steel can obtain better combination of strength and toughness after tempering.

3.4.3 The change of boron distribution during phase transformation

In order to gain the knowledge of the influence of phase transformation on the boron distribution, the Mn-Nb-B steel was used to study the relation between the isothermal ferrite and bainite transformation with the boron distribution^{29,30}. Figure 14 shows that boron distribution and the corresponding optical micrograph during ferrite transformation. Figure 14(a) demonstrates that there is no boron segregation along grain boundaries basically when the sample is direct quenched into ice brine after isothermal holding 15 minutes at 1200°C. The ferrite transformation is not starting but there is an obvious boron segregation on the original austenite grain boundaries when the sample is quenched into tin-bath furnace at 700°C for 5s as shown in Fig.14(b). As the holding time increases at 700°C, the ferrite formed along the grain boundary gradually, the metallographs and boron distribution of sample isothermally held for 120s are shown in Figs.14(c) and 14(d) respectively. As can be seen that apart from the continuous boron segregation on original austenite grain boundary, some high boron density ranges were visible along a part of the grain boundary (E F G), in another part of area the boron density is relatively lower than E F G ranges but at these positions the original grain boundary is invisible (H K). From the corresponding metallographs it is indicated the grain boundary noted by C, D is an original boundary and at E F G regions a bit of ferrite formed along the triple point of austenite grain boundaries. At H K region there are quite bigger growing ferrite block. Figures 14(e) and 14(f) show that in a different part of the same sample, in this region the ferrite formed earlier and they grew to touch

each other. These figures show that when the ferrite is growing the boron enrichment will decrease and disappear at last (E F). These results indicate that during the early stage of isothermal holding the boron atom would segregate quickly to the austenite grain boundaries in a non-equilibrium way. After the ferrite transformation started, the boron concentration in ferrite region is higher than any others. As the ferrite is growing the enrich phenomenon of boron in this region will decrease and disappear finally. The reason for this is that at the isothermal holding temperature the inverse diffusion rate of boron in ferrite is lower than the ferrite growing rate, the boron segregation existing at austenite grain boundaries can only disperse in the new ferrite region and can not catch the moving phase boundary and diffuse into austenite region. It also shows that the boron atoms segregated on austenite grain boundaries during quenching and isothermal holding process is in solution, which retards the ferrite formation along grain boundary. This phenomenon also evidences that the boron content is not a control factor of the proeutectoid ferrite transformation.

Fig. 14. boron distribution and the optical micrograph during ferrite transformation

It is also found that there is an obvious abnormal boron segregation at the moving phase boundaries between the original austenite and bainite block during the isothermal bainite transformation. The boron enrichment band along the boundaries between austenite and bainite will persist during the bainite interface is moving. The boron segregation will decrease and vanish at last when the transformation is completed and the bainite blocks meet

each other²⁹). **Figure 15** shows the boron distribution at the begin of the bainite transformation. There is an obvious boron segregation at the austenite grain boundaries and there is another obvious abnormal boron segregation at the moving boundary which is the boundary of new born bainite block just moving into the original austenite grain. This segregation is much heavier than that at the original grain boundaries, and it seems that this boron segregation will influence the bainite block growing process.

Fig. 15. boron distribution of bainitic group and austenite

3.5 Influence of boron on the aging hardening of low carbon bainitic steels

Due to the ϵ -Cu precipitation during aging, the bainitic steel with copper will be strengthened more over. There is a practical significance to study whether the boron has an influence on the ageing precipitation behavior. **Figure 16** shows the hardness change of 1[#] and 2[#] steels quenched into the saturated brine from 920°C and then aged at 520°C. It seems that boron increases the hardness slightly. The two peaks appearing at the curves of hardness changing with the aging time are proved to be caused by ϵ -Cu and Mo, Cr carbide aging precipitation according to Wang³¹). The hardness curves during aging of two steels with B or not are of similar characteristics indicated that at 520°C aging the boron has little impact on the ϵ -Cu and Mo, Cr carbide aging precipitation.

Fig. 16. hardness change of 1[#] and 2[#] steels aged at 520°C

3.6 Effect of boron on the weldability of low carbon steels

The properties of the HAZ (Heat affected zone) of low carbon bainitic steels have been studied by Tan³²), it was indicated that the combined addition of niobium and boron will guarantee to obtain a fine low carbon bainite with high toughness in HAZ. After different kinds of welding thermal cycle simulation, the HAZ structures have high strength and toughness, the FATT is below -40°C and the impact energy at room temperature is over 200J.

Fig. 17. The behavior of boron in ULCB steel during reheating at a rate of 100°C/s (a) 1000°C, (b) 1150°C, (c) 1350°C

The boron distribution during welding simulation has been investigated, it shows that when samples are quick heated at 100°C/s the boron constituent can not be dissolved when temperature is below 1000°C, meanwhile a few fine austenite grains already form and there is a boron segregation along these new grain boundaries. When heated to 1350°C, the boron segregation disappears (Fig. 17). If the heating rate increases to 500°C/s, the temperature at which new grains appears increases. The transformation from bainite to austenite can not be completed at 1350°C and new austenite grains can not fully meet each other, so a part of moving austenite grain boundaries still have boron segregation. It seems that the undissolved boron constituent can accelerate the austenite nucleation during reheating nevertheless they will hinder the grain boundaries moving. When the boron constituent are dissolved, there will be an obvious abnormal solute segregation at new moving grain boundaries, due to the solute dragging force the growing of austenite grain will be hindered. During subsequent cooling processing, a non-equilibrium boron segregation reappearing at grain boundaries can suppress the ferrite formation and hinder the growth of bainite, so the finer bainite structure can be obtained at HAZ.

4. Conclusions

- (1) Trace boron can enhance the hardenability of steels obviously and the effectiveness is determined by variation of non-equilibrium boron segregation at grain boundaries. If the boron segregates to grain boundary too quickly or too slowly during cooling process the boron effect on hardenability will not be the best. The heat history of steel will affect the behaviors of boron segregation markedly. The addition of Nb, Cu, Ti has a synergistic effect with boron and enhances

boron effect remarkably.

- (2) During high temperature deformation and recrystallization, there is a strong abnormal boron segregation on the moving new grain boundary, the nucleation and growing of new grains will be retarded. When Nb and Cu are added in the steel, the synergistic effect of Nb-B and Cu-B accelerates the Nb(C, N, B) precipitation and retards the recrystallization more over.
- (3) Trace boron can accelerate Nb(C, N) precipitation remarkably in steel with Nb and the precipitation start and finish times tend to be shorter. The nose temperature of PTT curve rises from 800°C to 850°C
- (4) Boron can reduce the start and finish temperatures of bainite transformation evidently and improve the mechanical properties of as rolled plate, and after tempering the boron steel will get better combination of strength and toughness. During tempering the lath-like microstructure of boron steel is more stable and the recovery process is slowed down. Boron makes the coarsening of austenite grains slow down during reheating in welding process, a finer bainite microstructure will be obtained at HAZ, and the toughness will be improved.
- (5) Boron has little impact on the ϵ -Cu and alloy carbides precipitation hardening during aging at low temperature.
- (6) The effect of boron in steel has relation to the different kinds of boron segregation behaviors. The addition of alloy elements has remarkable influence on the boron behavior. The study of relevant mechanism has made a marked progress, but there are a lot of questions need to be clarified.

REFERENCE

- 1) S. Wang and P. Kao: *China steel technical report.*, 6(1992), 1.
- 2) M. Krishnadev, G. Sojka and S. Banerji: *Trans. ASME.*, 103(1981), 207.
- 3) H. Tamehiro, M. Murata, R. Habu and M. Nagumo: *Trans. Iron Steel Inst. Jpn.*, 27(1987), 120.
- 4) H. Tamehiro, M. Murata R. Habu and M. Nagumo: *Trans. Iron Steel Inst. Jpn.*, 27(1987), 130.
- 5) J. Morral and T. Cameron: Boron in steel, ed. by S. Banerji and J. Morral, AIME, Warrendale, PA(1979), 235
- 6) R. Habu, M. Miyata, S. Sekino and S. Goda: *Trans. Iron Steel Inst. Jpn.*, 18(1978), 492.
- 7) S. Yang, S. Zhang, X. He, M. Chen and T. Ko: Microstructure LCS'94, ISIJ, Tokyo, (1994), 39.
- 8) X. Wang, G. Zhou S Yang and X. He: *J. Univ Sci & Techn, Beijing.*, 22(2000) 70.
- 9) X. Wang, S Yang, X. He, M. Wang and G. Zhou: *J. Iron steel Research*, 22(2000), 47.
- 10) C. Shang, X. Wang, X. He, S. Yang and Y. Yuan: *J. Univ Sci & Techn, Beijing.(English Edition)*, 8(2001), 224.
- 11) S. Yang, C. Shang, X. He, X. Wang and Y. Yuan: *J. Univ Sci & Techn, Beijing.(English Edition)*, 8(2001), 119.
- 12) S. Yang, X. Wang, C. Shang, X. He and Y. Yuan: *J. Univ Sci & Techn, Beijing.(English Edition)*, 8(2001), 214.
- 13) E. Czyryca, R. Link, R. Wong, D. Aylor, T. Montemarano and J. Gudas: *Naval engineers journal*, 102(1990), 63

- 14) L. Kvidahl: *Weld. J.*, **64**(1985), 42.
- 15) T. Montemarano, B. Sack, and J. Gudas: *Journal of ship production*, **8**(1986), 145.
- 16) Y. Ohmori and K. Yamanaka: Boron in steels(in Chinese), ed. by R. Zu and M. Guo, Metallurgy Industry Press, Beijing, (1985), 56.
- 17) X. He, Y. Chu and T.Ko: *Acta Metall Sinica.*, **19**(1983), A459
- 18) T. Koseki, K. Amano, M. Tanaka, C. Shiga and S. Ueda: Physical metallurgy of direct-quenched steels, ed. by K. Taylor , S. Thompson and F. Fletched, TMS, Warrendale, PA, (1993), 297.
- 19) X. He, S.Yang and T.Zhou: THERMEC' 97 Int. Conf. on Thermomechanical Processing of Steel and other Materials, ed. by T.Chandra and T.Sakai, TMS, Warrendale, PA, (1997), 443
- 20) G. Zhou: Doctor's dissertation, USTB , (2000)
- 21) X. He, M. Djhazi and J. Jonas: *Acta Metall. Mater.*, **39**(1991),225
- 22) T. Zhou: Master's dissertation, USTB, (1998)
- 23) S. Zhang, X. He and T. Ko: *J. mater. Sci.*, **29**(1994),2663
- 24) H. Cui, X. He and T. Ko: *Chinese Science Bulletin*, **40**(1995), 812
- 25) M Djahazi, X. He and J. Jonas: *Mater. Sci. Technol.* **8**(1992), 628.
- 26) S. Yang, X. He, M. Chen, Z. Dang and T. Ko: *Mater. Sci. Eng.*, **12**(1994), 49
- 27) G. Zhou, M. Wen, P. Li, X. He: *Acta Metall. Sin.(English letter).*, **13**(2000), 623.
- 28) X. Wang: Doctor's dissertation, USTB, (2000)
- 29) S. Yang: Doctor's dissertation, USTB, (1997)
- 30) S. Yang and X. He: *Acta Metall. Sinica.*, **35**(1999),23.
- 31) X. Wang, G. Zhou, S. Yang and X. He: *Acta Metall. Sinica.*, **36**(2000), 113.
- 32) D. Tan, S. Zhang, X. He, Z. Yu, T. Ko, Y. Chen and Y. Zhang: Microstructure LCS'94, ISIJ, Tokyo, (1994), 95.